Structural, electrical and photovoltaic characterization of Si nanocrystals embedded SiC matrix and Si nanocrystals/c-Si heterojunction devices

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Abstract

Thin films of Si nanocrystals (Si NCs) embedded in a silicon carbide (SiC) matrix (Si-NC:SiC) were prepared by alternating deposition of Si-rich silicon carbide (Si_{1-x}C_{x}) and near-stoichiometric SiC multilayers (Si_{1-x}C_{x}/SiC) using magnetron cosputtering followed by a post-deposition anneal. Transmission electron microscopy and Raman spectroscopy revealed that the Si NCs were clearly established, with sizes in the range of 3–5 nm. Optical studies showed an increase in the optical band gap after annealing from ~1.4 eV (as-deposited) to ~2.0 eV (annealed at 1100 °C). P-type Si-NC:SiC/n-type crystalline silicon (c-Si) heterojunction (HJ) devices were fabricated and their electrical and photovoltaic properties were characterized. The diode showed a good rectification ratio of $1.0 \times 10^4$ at the bias voltage of ±0.1 V at 298 K. The diode ideality factor and junction built-in potential deduced from current–voltage and capacitance–voltage plots are ~1.24 and 0.72 V, respectively. Illuminated $I–V$ properties showed that the 1-sun open-circuit voltage, short-circuit current density and fill factor of a typical HJ solar cell were 463 mV, 19 mA/cm² and 53%, respectively. The external quantum efficiency and internal quantum efficiency showed a higher blue response than that of a conventional c-Si homojunction solar cell. Factors limiting the cell’s performance are discussed.

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1. Introduction

Over the past decade, materials consisting of silicon nanocrystals (Si NCs) embedded in a dielectric matrix have been the subject of intense research in the field of optoelectronics [1,2]. Low dimensional Si NCs efficiently emit light even at room temperature due to quantum confinement effects, thus potentially allowing optoelectronic functions to be incorporated into Si integrated circuit technology. Recently, Si NCs have also attracted interest in relation to fabrication of “all-Si” tandem cells for third generation photovoltaics [3,4]. Si NC materials may allow the fabrication of higher band gap solar cells that can be used as tandem elements on top of a conventional Si cell. It is possible that band gap for tandem elements and sufficient carrier mobility can be achieved by optimizing Si-NC size, spacing and barrier height between matrix and Si NCs [4,5]. Although Si NCs embedded in a silicon oxide or silicon nitride have been widely investigated [6,7], Si NCs in a SiC matrix (Si-NC:SiC) are of particular interest as the low barrier height of SiC relative to SiO₂ and Si₃N₄ is conducive to carrier transport [8]. However, experimental investigations on Si-NC:SiC films are few [9] and the electrical and photovoltaic properties of devices based on these films are not reported yet.

In this work, Si-NC:SiC films were prepared by alternating deposition of substoichiometric and near-stoichiometric silicon carbide multilayers (Si_{1-x}C_x/SiC) by magnetron cosputtering followed by a post-deposition anneal. The use of Si_{1-x}C_x/SiC multilayers instead of a single Si_{1-x}C_x layer is expected to give better control over
Si-NC size as the Si NCs are constrained by the thickness of the Si-rich layer. This was found to be the case for Si NCs in the SiO2 and Si3N4 matrix analogues studied in Refs. [3,6]. We present the results of structural studies by transmission electron microscopy (TEM) and Raman spectroscopy. The chemical composition depth profile has been determined by X-ray photoelectron spectroscopy (XPS) and secondary ion mass spectroscopy (SIMS). The optical band gap energy ($E_{\text{gopt}}$) at room temperature was determined by the Tauc’s equation. Heterojunction (HJ) devices of $p$-type Si-NC:SiC on $n$-type crystalline silicon substrates [(p)Si-NC:SiC/(n)-Si] were fabricated. The study of the (p) Si-NC:SiC/(n) $c$-Si HJs is useful to evaluate the electrical material quality of the Si-NC:SiC films, and to investigate the design parameters required for their application in photovoltaic devices. Additionally, it is also useful to build on the understanding of properties for this kind of HJs because it is of considerable interest as wide band gap emitters or window regions in bipolar transistors [10] and photodetectors [11]. The electrical properties of the devices were characterized by dark current–voltage ($I$–$V$) and capacitance–voltage ($C$–$V$) measurements. The photovoltaic properties were evaluated by illuminated $I$–$V$, quasi-steady-state open-circuit voltage (Suns-$V_{\text{oc}}$), and quantum efficiency measurements. Factors limiting the cell’s performance are discussed.

2. Experimental procedure

Si wafer and quartz plate substrates were used simultaneously for different measurement purposes. The Si wafer substrates were used for structural and electrical performance measurements, and the quartz plates were used for optical measurements. The Si and quartz substrates were cleaned by standard wet chemistry cleaning procedures. The Si wafers were additionally dipped in 5% HF solution for 10 s to remove the surface native oxide before being immediately loaded into the sputtering system vacuum chamber. The amorphous Si$_{1-x}$C$_x$/SiC ($x$=0.1–0.15) multilayers were deposited by magnetron cosputtering from Si and SiC targets using a computer-controlled AJA ATC-2200 sputtering system. The SiC target was boron-doped, with the intention that boron would be incorporated in the multilayer films. The thickness of individual layers was typical $\sim$6 nm (Si$_{1-x}$C$_x$) and $\sim$2.5 nm (SiC) and the total emitter thickness was $\sim$160 nm. Prior to sputtering, the chamber was evacuated down to a pressure of $\sim$5 x 10$^{-7}$ Torr, subsequently the chamber was filled with Ar gas to a working gas pressure of 2 mTorr. The argon flow was maintained at 20 sccm during deposition. All the samples were not intentionally heated during the deposition process. Following deposition the samples were annealed at 1100 °C for 9 min in a conventional furnace in a N$_2$ ambient.

The structures of the Si-NC:SiC films were investigated by cross-sectional TEM (JEOL-3000F, operating at 300 kV). Depth profiling of the films’ chemical composition was conducted by XPS (Fisons ESCALAB 220i-XL) and SIMS. The SIMS analysis was done by a 25 kV Bi$^+$ TOF-SIMS IV (time of flight SIMS) system, ION-TOF with sputtering of 0.5 kV Cs$^+$ ions at a 45° incident angle. The beam currents were 1 pA and 30 nA for the analysis gun and sputtering gun, respectively. Raman spectra were measured by micro-Raman spectroscopy (Renishaw, RM2000) in a backscattering configuration, with a 50 x optical microscope objective. The laser beam was provided by an Ar ion laser with a wavelength of 514.4 nm. A double-beam UV/visible/IR spectrophotometer (Varian Cary 5G) and an attached integrating sphere (Labsphere, RSA-CA-50) were used to measure transmission and reflection spectra. The dark $I$–$V$ characteristics of the diodes were measured by an automatic measuring system comprised of a DC voltage–current source/monitor (Advantest, TR6143) and a computer. The measured devices were held on a vacuum chuck whose temperature was controlled by a temperature controller. A copper casing was used to cover the sample holder to prevent environmental light from influencing the recorded data. $C$–$V$ measurements were performed by an impedance analyzer (HP4194) at a frequency of 100 kHz. The spectral response was determined through measurement of the spectral dependence of the external quantum efficiency (EQE) and internal quantum efficiency (IQE) over the wavelength range of 300–1200 nm. The 1-sun $I$–$V$ curves of the solar cells were measured under approximated standard test conditions (AM1.5G spectrum, 100 mW/cm$^2$, 298 K). The pseudo $I$–$V$ curves without the influence of the series resistance were extracted by means of the quasi-steady-state open-circuit voltage measurements (abbreviated “Suns-$V_{\text{oc}}$” in this work).

3. Results and discussion

3.1. Film characterization

3.1.1. Structural and compositional properties

The presence of Si NCs in the annealed films was confirmed by TEM observations. Fig. 1 shows a typical cross-sectional TEM view of an as-deposited sample. In the lower left corner of Fig. 1 is an inset of the high resolution TEM image of the sample annealed at 1100 °C for 9 min. The as-deposited film contains two uniform amorphous phases with a clear layered structure due to the density difference between Si$_{1-x}$C$_x$ and SiC materials. After annealing, we can clearly see lattice fringes (see the inset) that are dispersed in amorphous matrix, indicating the formation of NCs after annealing. A measurement of the lattice fringes of the high-resolution TEM image demonstrates the NCs are composed of Si atoms because the lattice fringes, at 3.1 Å, correspond to Si {111} lattice planes. The Si-NC size obtained from the TEM micrographs is in the range of 3–5 nm. Fig. 2 shows a typical Raman spectrum of the Si-NC:SiC film on a quartz substrate formed by annealing at 1100 °C [curve (a)]. The
Raman spectra of the as-deposited sample [curve (b)] and c-Si [curve (c)] are also included for comparison. For the as-deposited sample [curve (b)], a broad peak around 470 cm$^{-1}$ is attributed to the stretching vibration mode of the Si–Si bonds due to the transverse optical (TO) mode of the amorphous silicon network [12]. As a result, we conclude that the as-deposited films are completely amorphous, which is consistent with the TEM results.

However, a significant change is observed in the Raman spectra of the annealed samples. An intense peak at around \( \sim 511 \text{ cm}^{-1} \) is assigned to the contribution from Si NCs [13]. Compared to the Raman peak of the c-Si substrate at \( \sim 520 \text{ cm}^{-1} \) [see curve (c) in Fig. 2], one of the remarkable features in the Raman spectra of the annealed sample is the appearance of a large frequency downshift (\( \Delta \omega = 9 \text{ cm}^{-1} \)), which is considered to be caused by grain size related effects in small-grained nanostructures and compressive stress in the films [14,15]. A 9 cm$^{-1}$ Raman peak shift to lower wave numbers may be caused by Si NCs with a \( \sim 3\)-nm size according to Faraci et al. [13,16] who investigated in the dependence of Si-NC size on Raman measurements. In addition to the Raman peak from Si NCs, we also observed a broad peak at \( \sim 935 \text{ cm}^{-1} \). Considering that the amorphous SiC vibrational density of states in Raman spectra is up to \( \sim 900 \text{ cm}^{-1} \) [17] and that the maximum optical phonon energy of any of the crystalline polytypes of SiC is 972 cm$^{-1}$ [18], the Raman band at \( \sim 935 \text{ cm}^{-1} \) in curve (a) of Fig. 2 can be explained by changes in the SiC bonding.

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**Fig. 1.** Typical cross-sectional TEM image of as-deposited multilayers on a Si substrate. The inset is a high resolution TEM image of the sample annealed at 1100°C for 9 min, indicating the Si NCs formed after annealing. The visible Si NCs in the inset are highlighted by the white circles.

**Fig. 2.** Raman spectra of the: (a) annealed film, (b) as-deposited film and (c) crystalline silicon. The origins of major peaks are indicated on the appropriate curves.
states, from amorphous to nanocrystalline, during annealing. This was confirmed by XRD measurements [19], showing the formation of Si NCs is accompanied by the presence of $\beta$-SiC in the annealed films. The fraction of $\beta$-SiC depends on the sample stoichiometry [19]. Notably, the Raman band at $\sim 935$ cm$^{-1}$ has previously been attributed to the longitudinal optical (LO) phonon band of microcrystalline SiC [20].

XPS analysis of the film was performed to determine the chemical composition depth profile. Fig. 3 shows a typical result of the XPS depth profile of Si, C and O for the annealed film on a Si substrate (note that the quantity of doped boron in the films is well below the detection limit). We note that the concentration of C and Si in the annealed film is virtually constant. This is most probably due to interdiffusion of C and Si atoms through the Si$_{1-x}$C$_x$/SiC multilayers during annealing. The increased carbon and oxygen concentration at the surface is due to surface contamination arising from exposure to air. At larger depths approaching the film-substrate interface it can be seen that the carbon fraction decreases sharply, whereas the silicon increases up to 100%, corresponding to the Si substrate. Oxygen in the films was most likely caused by residual atmospheric gases in the chamber of the sputtering system.

3.1.2. Optical band gap
The optical band gaps ($E_{\text{gopt}}$) were determined from optical absorption measurements of the films that were deposited on quartz substrate. $E_{\text{gopt}}$ can be extracted using the following formula:

$$(\alpha h\nu)^{\gamma} = B(h\nu - E_{\text{gopt}}),$$

where $h$ is Planck’s constant, $\nu$ is the frequency of the radiation, $\alpha$ is the optical absorption coefficient, and $\gamma$ is a value dependent on the $E_{\text{gopt}}$ behavior (direct $E_{\text{gopt}}$ $\gamma = 2$, and indirect $E_{\text{gopt}}$ $\gamma = 1/2$) [21]. $B$ is the edge width parameter, which may be used as a measure for the structural disorder of films [22]. The optical absorption coefficient can be deduced by transmission $T$ and reflection $R$ data. As a first approximation, $\alpha$ is related to $T$ and $R$ as follows [23]:

$$T = (1 - R) \exp(-\alpha t),$$

$$\alpha = \frac{1}{t} \ln \left( \frac{1 - R}{T} \right) \text{ cm}^{-1},$$

where $t$ is the thickness of the film. Eq. (1) has been used by some authors to extract the band gap for different NC materials [21,24–27] although there exists an argument for suitability of Tauc equation for NC systems due to the NC materials having a distribution of the band gaps [28].

Fig. 4 shows Tauc’s plot of $(\alpha h\nu)^{1/2}$ versus photon energy $h\nu$ of a sample before and after annealing, where $\gamma = 1/2$ was used since a phonon-assisted transition still dominates the optical processes even for a small Si-NC size [21,24,29]. $E_{\text{gopt}}$ is evaluated by extrapolating the linear part of the $(\alpha h\nu)^{1/2}$ versus incident photon energy plot to intercept the energy axis $(\alpha h\nu = 0)$. It is clearly observed that the absorption edge can be linearly fitted by Eq. (1) with $\gamma = 1/2$, and exhibits a shift to higher photon energies upon annealing. The $E_{\text{gopt}}$ increases from $\sim 1.4$ eV (as-deposited) to $\sim 2.0$ eV (annealed at 1100 °C). However, it should be noted that a detail investigation of optical properties of Si-NC:SiC [30] for different annealing temperatures and C/Si ratio shows that Tauc’s plot may overestimate the $E_{\text{gopt}}$ because the band gap caused by Si
NCs is likely masked by absorption features of the bonded SiC in the matrix.

3.2. Si-NC:SiC/c-Si HJ device

3.2.1. Device fabrication

P-type Si-NC:SiC/n-type c-Si HJ devices were fabricated to evaluate the suitability of Si-NC:SiC for photovoltaics applications. The device substrates are n-type silicon wafers (100)-oriented, with a resistivity of 2–9 $\Omega \cdot cm$, corresponding to a doping density of (0.4–2.5) $\times 10^{15}$ cm$^{-3}$. The as-deposited Si$_{1-x}$C$_x$/SiC multilayers have a total thickness of $\sim$160 nm. After 1100 °C annealing, the film’s resistivity is in the range $10^{-1}$–$10^{-2}$ $\Omega \cdot cm$, measured by using the four-point probe method. This low resistivity can be associated to the activation of boron dopants and Si-NC formation due to annealing [31,32]. Ohmic contacts fully covering the back surface of the n-type wafers were fabricated by vacuum evaporation of Ti ($\sim$30 nm) followed by Al ($\sim$1.0 $\mu$m). The front electrode consists of an Al metal grid (thickness $\sim$0.8 $\mu$m) formed by evaporating Al through a Si shadow mask. A thermal baking at 400 °C for 40 min in a N$_2$ atmosphere was performed to improve the quality of the Ohmic contacts. A final cell area of 1.0 $\times$ 1.0 cm$^2$ was defined by a Q-switched pulsed Nd:YAG laser scribing. No anti-reflection coating and texturing were used in the fabricated cells. Fig. 5 shows a schematic diagram of a (p) Si-NC:SiC/(n) c-Si HJ solar cell.

During annealing, impurity atoms from the deposited layer may diffuse into the substrate and move the junction away from the Si-NC:SiC/c-Si interface [33,34]. In this case, a (p)Si-NC:SiC/(p)c-Si/(n)c-Si structure can be formed [35]. To clarify the effect of annealing on impurity redistribution of the devices, SIMS measurements were performed to investigate the depth profile of the dopant (boron) across the HJ interface. Fig. 6 shows a comparison of the SIMS relative boron profiles for as-deposited and annealed (1100 °C for 9 min) HJ devices. The Si profile is also included as a reference. Due to a shrinkage effect during annealing, the annealed film has a slightly shorter sputtering time from the device surface to the Si substrate. For the as-deposited device, the Si concentration exhibits an oscillatory distribution throughout the film. Twenty peaks and valleys can be resolved, corresponding to the layered structure disappears and Si concentration becomes uniform, consistent with previous XPS results (see Fig. 3). In contrast to Si, the boron profiles remains the nearly same throughout the entire film. At the HJ interface it can be seen that the boron concentration decreases sharply, whereas that of silicon increases up to a constant level, corresponding to the c-Si substrate. Comparing the boron profiles at the HJ interface before and after annealing, the similar SIMS boron profiles indicate that short annealing time does not have a significant effect on the junction position although the relative boron profiles do not rule out completely the possibility of boron diffusion from the Si-NC:SiC film into the Si wafer to create a shallow buried homojunction.

3.2.2. $I$–$V$ and $C$–$V$ characteristics

Fig. 7 shows the dark $I$–$V$ characteristics of a (p) Si-NC:SiC/(n) c-Si HJ diode measured at room temperature in both polarities. For analysis purpose, the $I$–$V$ curves are plotted on two scales: (a) the semi-logarithmic plot and (b) the log–log plot. As can be seen, the diode shows a good rectification ratio of $1.0 \times 10^5$ at $\pm 1.0$ V. If the forward-biased curve is fitted to a standard diode equation, we find that the ideality factor of the HJ diode is $\sim$1.24 in intermediate bias voltages ($V \sim 0.1–0.4$ V). From the log–log plot [Fig. 7(b)], three distinctly different conductive regions can be identified (labeled region I, II and III): (i) a low forward bias segment (region I, $V < \sim 0.1$ V), where the current increases linearly with applied voltage, (ii) an intermediate voltage segment (region II, $V \sim 0.1–0.4$ V), where the current increases exponentially with the applied voltage, and (iii) a high voltage segment (region III, $V > \sim 0.4$ V), where the current
increases at a slower rate and deviates from exponential behavior. The different conductive regions indicate a corresponding change in the carrier transport mechanism. A detailed analysis of dominant carrier transport mechanisms in each region has been performed by means of temperature-dependent current versus voltage ($I–V–T$) measurements (see Ref. [36] for details). In brief, the current in the low bias region is controlled by shunt resistance, whereas space-charge limited current (SCLC) conduction becomes dominant in the high bias region. At intermediate bias voltages, junction interface recombination is the main transport mechanism.

$C–V$ measurements were performed on the diode from which the majority carrier barrier height of the HJ and the doping concentration in the base layer can be extracted. For an anisotropic HJ, assuming a uniform doping distribution in the device, the $1/C^2$ versus the applied voltage $V$ can be expressed by Anderson’s model [37]

$$\frac{1}{C^2} = \frac{2}{qA} \left( \varepsilon_{Si} N_{D, Si} + \varepsilon_{NC} N_{I, NC} \right) (V_{bi} - V),$$

(4)

where $q$ is the electronic charge, $\varepsilon_0$ is the vacuum permittivity, $\varepsilon_{Si}$ and $\varepsilon_{NC}$ are the dielectric constants of Si and Si-NC:SiC, respectively. $N_{D, Si}$ is the donor doping concentration in the n-type Si substrate, $N_{I, NC}$ is the effective density of acceptor-states in Si-NC:SiC film, $A$ is the junction area, and $V_{bi}$ is the built-in potential. The $1/C^2$ is expected to be linear function of the reverse bias voltage, and $V_{bi}$ can be obtained from intercept point of the linear curve and voltage axis. Fig. 8 shows the measured $1/C^2–V$ characteristics of a typical Si-NC:SiC/c-Si HJ diode at a frequency of 100 kHz at 298 K. This plot reveals a good linear relationship in the wide voltage range from $-5.0$ to $0.1$ V, indicating that the abrupt HJ theory is applicable to the devices. The $1/C^2–V$ intercept of $\sim 0.72$ V (obtained by extrapolating the linear fit to the voltage axis) is essentially equal to the barrier height within the Si substrate. Considering that the depletion region is predominantly on the Si side of the junction and $N_{I, NC} > N_{D, Si}$, the donor doping concentration in the base layer can be calculated from the slop of the $1/C^2$ versus bias voltage plot using Eq. (4). The slope of the curve in Fig. 8 gives an $N_{D, Si}$ value of $2.2 \times 10^{15} \text{cm}^{-3}$, which agrees well with the resistivity of the n-type Si substrates used in the HJ devices.
3.2.3. Photovoltaic properties

Fig. 9 shows the 1-sun $I-V$ curve of a typical ($p$) Si-NC:SiC/($n$) c-Si HJ solar cell measured under the standard AM1.5G spectrum with an illumination level of 100 mW/cm$^2$ at 298 K. The cell has a $V_{oc}$ of 463 mV, a $J_{sc}$ of 19 mA/cm$^2$, a fill factor of 53%, and an energy conversion efficiency of 4.66%. The inset is a pseudo $I-V$ curve without the influence of the series resistance extracted from the $Suns-V_{oc}$ curve. Without the influence of the series resistance extracted from the $Suns-V_{oc}$ method [38,39]. Examining the illuminated $I-V$ and pseudo $I-V$ curves, there is a significant reduction in the real fill factor (53%) compared to the pseudo-FF (77%). It is apparent that the reduction in the real fill factor is mainly due to the high series resistance, which limits both $J_{MP}$ and $V_{MP}$ (current density and voltage at maximum power point) of this device. A series resistance $R_{s\text{light}}$ of 4.72 $\Omega$cm$^2$ was extracted by a fit to the illuminated $I-V$ curve. For the low $V_{oc}$ of HJ solar cells, in principle, the main limiting factor is recombination in junction interface, recombination in back contact, and bulk recombination. According to the dark $I-V$ analysis in Section 3.2.2, the low $V_{oc}$ (463 mV) should be mainly limited by interface quality because recombination through interface states is the dominant current transport mechanism at intermediate bias voltages.

Fig. 10 shows the reflection ($R$), EQE and IQE of a typical ($p$) Si-NC:SiC/($n$) c-Si HJ solar cell. For the blue wavelength response, the IQE is $\sim$35% at 400 nm, which appears to be higher than that of conventional c-Si homojunction solar cells [40]. This may be due to either the wide band-gap Si-NC:SiC window layer or some contribution to absorption by possibly the Si NCs in the SiC matrix, although this is far from conclusive. Integration of the measured EQE and IQE curves for 1-sun AM1.5 illumination gives $J_{sc} = 19.46$ mA/cm$^2$ (EQE) and $J_{sc} = 30.86$ mA/cm$^2$ (IQE), respectively. The former is in good agreement with $J_{sc} = 19$ mA/cm$^2$ determined from the illuminated $I-V$ curve. The large $J_{sc}$ difference between the EQE and the IQE is due to a high front surface reflection. The relatively poor full-wavelength response of the IQE (peak at $\sim$89%) suggests a large collection loss of photogenerated carriers in the cells. It is postulated that defects at the junction interface are formed during high temperature annealing, due to the large difference in thermal expansion coefficient between Si and SiC (8%). These defects at the interface can attract space charges and/or act as recombination centers, which would result in a decrease in photocurrent through the junction and a subsequent reduction in $V_{oc}$.

4. Conclusions

We have fabricated Si-NC:SiC films and ($p$) Si-NC:SiC/($n$) c-Si HJ devices by magnetron sputtering of Si$_{1-x}$C$_{x}$/SiC multilayers followed by a post-deposition anneal. The structural, compositional and optical properties were investigated by TEM, Raman spectroscopy, XPS and optical transmission and reflectance measurements. TEM and Raman spectroscopy investigations showed evidence of Si NCs embedded in a SiC matrix, with sizes in the range of 3–5 nm. XPS and SIMS analysis showed uniform film chemical compositions for the annealed samples and a sharp interface at the HJ. Optical studies showed an increase in the optical band gap after films were annealed, from $\sim$1.4 eV (as-deposited) to $\sim$2.0 eV (annealed at 1100 °C). The ($p$) Si-NC:SiC/($n$) c-Si diode has a good rectification ratio of $10^4$ at the bias voltage of $\pm$1.0 V at 298 K. The diode ideality factor and junction built-in potential are $\sim$1.24 and 0.72 V, respectively. Illuminated $I-V$ measurements showed that the 1-sun open-circuit voltage, short-circuit current density and fill factor of a
typical HJ solar cell are 463 mV, 19 mA/cm² and 53%, respectively. The EQE and IQE measurements showed a higher blue response than that of a conventional c-Si homojunction solar cell. For photovoltaic properties, most of the absorption in this device originates from the Si substrate and hence absorption contribution from the Si-NCs cannot be established. Nevertheless, the demonstration of a Si-NC:SiC/c-Si HJ photovoltaic device is an encouraging step towards realization of an all-Si tandem cell based on Si-NC materials.

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References